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J. Gayda, T.P. Gabb, R.V. Miner,
and G.R. Halford
Lewis Research Center
Cleveland, Ohio

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BITHERMAL LOW-CYCLE FATIGUE BEHAVIOR OF A NiCoCrAlY-COATED SINGLE CRYSTAL SUPERALLOY

J. Gayda, T.P. Gabb, R.V. Miner, and G.R. Halford
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio 44135

SUMMARY

Specimens of a single crystal superalloy, PWA 1480, both bare and coated with a NiCoCrAlY alloy, PWA 276, were tested in low-cycle fatigue at 650 and 1050 °C, and in "bithermal" thermomechanical fatigue tests. In the two bithermal test types, tensile strain was imposed at one of the two temperatures and reversed in compression at the other. In the high-strain regime, lives for both bithermal test types approached that for the 650 °C isothermal test on an inelastic strain basis, all being controlled by the low ductility of the superalloy at 650 °C. In the low-strain regime, coating cracking reduced life in the 650 °C isothermal test. The bithermal test imposing tension at 650 °C, termed "out-of-phase," also produced rapid surface cracking, but in both coated and bare specimens. Increased crack growth rates also occurred for the out-of-phase test. Increased lives in vacuum suggested that there is a large environmental contribution to damage in the out-of-phase test due to the 1050 °C exposure followed by tensile straining at the low temperature.

INTRODUCTION

Gas turbine engine components are subject to complex local stress-strain-temperature-time cycles during operation. In addition to mechanical loads imposed centripitally and by gas impingement, thermal strains are imposed by temperature gradients and thermal expansion differences between joined materials. Damage from such cyclic loading is termed thermomechanical fatigue, or TMF. Advanced turbine engine blades are Ni-base superalloys cast as single crystals and coated in various ways with a layer which is basically more Al-rich for protection from the environment. Relative to the base superalloy, these coatings have greater coefficients of thermal expansion, higher elastic moduli than the [001] lengthwise axis of a single crystal turbine blade, lower flow stress, particularly at high temperature, good ductility at high temperature, but very low ductility at temperatures below about 700 °C (ref. 1). Cracking of the coating is promoted in areas of a component where it is cycled into tension at low temperatures and these cracks may then propagate into the superalloy below (refs. 2 and 3). The performance of a particular coating is further reduced if its coefficient of thermal expansion is greater than that of the superalloy, since it is placed in tension simply upon cooling from high temperatures (ref. 3).

The eventual aim of this research is to model failure of coated single crystal superalloy laboratory specimens during generalized TMF cycling by treating the specimen as a composite system and using behavior established for both the superalloy and bulk coating alloy. A system with the overlay type of coating has been chosen rather than the diffused-in type to simplify the problem. At least as they go into service, the overlay coatings have a uniform

composition and microstructure through most of their thickness. Further, they may be produced in bulk form by processes such as low pressure plasma spraying and have microstructures closely resembling those of the thin coatings deposited on real hardware.

The system chosen for study, Ni-base single crystal superalloy PWA 1480 and NiCoCrAlY low-pressure-plasma-sprayed overlay coating PWA 276, was developed by Pratt and Whitney Aircraft. The mechanical behavior of PWA 1480 crystals from the same lot studied herein have been described previously for isothermal monotonic (ref. 4) and creep-fatigue loading (ref. 5). Comparison of several studies of the monotonic loading behavior shows there is considerable variability among different lots of PWA 1480 (ref. 4). The bulk coating alloy PWA 276 produced as thick plates by the same low-pressure-plasma-spray process has also been studied in high temperature monotonic (refs. 6 and 7) and fatigue tests (ref. 8). DeLuca and Cowles (ref. 9) have studied another heat of PWA 1480 single crystals with a diffusion-aluminide coating. Both fatigue crack initiation and propagation rate tests were conducted using TMF cycles in which mechanical strain and temperature were varied simultaneously.

The present work describes the life and failure mechanisms of coated and bare crystals of PWA 1480 in isothermal fatigue at 650 and 1050 °C and "bithermal" TMF cycles involving mechanical straining only at these two temperature ends of the cycles. The temperature 650 °C represents about the upper limit of low temperature behavior in both materials characterized by relatively constant high strength and low ductility. While 1050 °C is about the upper limit of allowable temperature in turbine engine blades and is in the regime where the coating is extremely weak and ductile. The bithermal test was proposed earlier by one of the present authors (ref. 10). Experimentally, it simplifies analysis and control since the mechanical and thermal strains are interposed rather than superimposed. Also, this test offers the possibility of easier understanding of TMF since deformation, at least in the superalloy, occurs only at the two temperature limits. This allows comparison and possible unification with isothermal behavior at the two temperatures. The qualification above is presented because stresses during heating and cooling due to thermal mismatch likely produce some inelastic deformation in the coating at high temperatures before the application of external load.

MATERIALS AND PROCEDURES

Materials

The single crystal superalloy PWA 1480 and coating PWA 276 studied herein were developed by Pratt and Whitney Aircraft. PWA 1480 has the following nominal composition in weight percent: 10Cr, 5Al, 1.5Ti, 12Ta, 4W, 5Co, balance Ni. The alloy contains about 65 vol % of the γ' phase, but no carbides or borides.

The crystals were cast by a commercial vendor as round bars about 21 mm in diameter and 140 mm long. They were solution treated for 4 hr at 1290 °C before machining. Bars having [001] within 7° of the center line were selected for fatigue specimens. These had a 4.8 mm diameter by 15 mm long reduced section. After machining most specimens were coated by low pressure plasma spraying. The PWA 276 coating composition in weight percent was 20.3Co,

17.3Cr, 13.6Al, 0.5Y, and balance Ni. The coating thickness was about 0.13 mm. Both coated and uncoated specimens were given the coating cycle heat treatment of 1080 °C for 4 hr and aged at 870 °C for 32 hr.

The microstructure of the superalloy and coating are represented in figure 1. It may be seen that there were several types of nonuniformities in the superalloy. Interdendritic porosity is normal in cast single crystal superalloys, and those studied herein contained about 0.3 vol %. The pore diameter averaged 7 μm with a standard deviation of 5 μm. The interdentritic areas also contained undissolved γ' eutectic nodules a few micrometers to tens of micrometers in diameter. These occupied 1 to 2 vol % of the alloy. Elsewhere, the γ' was present as cuboids about 0.6 μm on edge. The PWA 276 alloy is about 50 vol % Ni-based solid solution and 50 vol % NiAl-based intermetallic compound. The low-pressure-plasma-sprayed coating had a grain size of about 1.5 μm and contained 1 to 2 vol % of pores averaging about 20 μm in diameter. The average surface roughness was 8 μm.

Test Procedures

All fatigue testing was done on 90 kN, servohydraulic, closed-loop test machines, one of which was equipped with a diffusion pumped vacuum chamber. Both machines were equipped with a 5 kW radio frequency induction generator for specimen heating. Closed-loop temperature control was employed in all tests, however, two methods of temperature measurement were used. For tests of coated specimens in air an infrared pyrometer was used. However, for tests of bare specimens in air and all tests in vacuum, emissivity was found to vary significantly with time and a Type K thermocouple was employed. The junction was pressed against, rather than welded to, the specimen by wrapping the wires around the specimen and applying a slight tension. The thermocouple was calibrated and periodically re-checked with a disappearing filament optical pyrometer, which is relatively insensitive to changes in emissivity. Strain was measured using an axial extensometer with a 12.5 mm gauge length. A dual pen strip chart was used to record strain/time and load/time data continually, while an x-y recorder was used to record load/strain hysteresis loops.

Isothermal fatigue tests were conducted at 650 and 1050 °C under total mechanical strain control at a frequency of 0.1 Hz. A digital function generator was employed to produce the sinusoidal control waveform with an R ratio of -1 (minimum/maximum strain).

TMF behavior was studied in a simplified bithermal cycle between 650 and 1050 °C. Specimens were strained at one temperature, unloaded, changed to the other temperature, and then strained in the opposite direction to produce a completely reversed strain cycle. Figure 2 shows the stress-strain hysteresis loop for what is termed an "out-of-phase" cycle in which the tensile and compressive strains are imposed at the lower and higher temperatures, respectively. Tensile strain and high temperature coincide in the "in-phase" cycle. A 16 bit computer, equipped with dual digital/analog converters was used to generate the control waveforms for load and temperature. Total cycle time was about 120 sec, of which 100 to 110 sec elapsed while changing and stabilizing temperature. The remainder of the time, 10 to 20 sec, was spent applying mechanical loads, approximating the 0.1 Hz frequency of the isothermal tests.

For large-strain, short-life bithermal tests, equal tensile and compressive inelastic strain ranges, $\Delta\varepsilon_{in}$, at the two test temperatures may be produced by fixing the endpoint sums of the total mechanical, $\Delta\varepsilon_t$, and thermal, $\Delta\varepsilon_{th}$, strains (fig. 2). Under this condition the material will rapidly equilibrate the tensile and compressive $\Delta\varepsilon_{in}$. A constant $\Delta\varepsilon_t$ test results if the temperature endpoints, and thus $\Delta\varepsilon_{th}$, are held constant. However, for small-strain, long-life tests, choosing the endpoints poses problems. When $\Delta\varepsilon_{in} \leq 10^{-4}$ it cannot be resolved, and when still lower, the test may be considered essentially elastic. Unlike the self-equilibrating large-strain tests, there are infinite combinations of endpoints which would produce essentially elastic tests, all with different maximum tensile and compressive stresses. These tests would have the equivalent of what would be termed different mean stresses in an isothermal test. The self-equilibrating large-strain tests do not. Tests with various endpoints representing reasonable extrapolations from higher strain range tests indicated no significant differences in life.

Tests in the low-strain regime were also conducted in load control with the tensile/compressive load ratio held the same as in strain-controlled tests at the lower ranges. The rationale was twofold. First, since loading is nearly elastic, the constant tensile/compressive load ratio should yield a cycle nearly "balanced" in terms of the minute $\Delta\varepsilon_{in}$ as desired in a strained-controlled test. But second, the stability and accuracy of the test improves. When the sum of the mechanical and thermal strains are fixed, fluctuations in the endpoint temperatures result in small mechanical strain fluctuations, and in low-strain tests these fluctuations become of the magnitude of the small $\Delta\varepsilon_{in}$. Control on the maximum tensile and compressive loads is not similarly affected. Out-of-phase tests in vacuum were conducted with tensile/compressive load limits identical to tests run in air. The vacuum was about 10^{-6} torr.

RESULTS

Cyclic Stress-Strain Behavior

Figure 3 compares the cyclic stress response of the PWA 1480 single crystals in isothermal and bithermal tests and that of the bulk PWA 276 coating alloy determined previously in isothermal tests (ref. 8). The comparison is made on the basis of the maximum absolute value of stress at 1050 °C for a given $\Delta\varepsilon_{in}$ in figure 3(a) and similarly for 650 °C in figure 3(b). The stresses measured at half life are used. The stress response of the single crystals as a function of the number of cycles was very stable for the 650 °C isothermal and bithermal tests, but about 5 to 10 percent softening occurred in the 1050 °C isothermal tests.

It may be seen that the strength of the coating is about one tenth that of the superalloy at 1050 °C, and at 650 °C it is about one third. Further, the area of a 0.13 mm thick coating on a 4.8 mm diameter substrate is about one tenth of that of the substrate. The stresses shown for the 1050 °C tests of the coated single crystals have been calculated assuming that the coating carried no significant load. This assumption appears justified by the agreement between the data for the coated and bare single crystals. This was not true at 650 °C. For the coated specimens, small corrections have been applied

to the stresses and inelastic strains in the superalloy for the 650 °C isothermal tests. As described in the appendix, it was possible to estimate the load borne by the coating with knowledge of the 650 °C cyclic stress-strain behavior of the coating and its elastic modulus (ref. 8). The load borne by the coating increases the apparent $\Delta\varepsilon_{in}$ at 650 °C, however in the bithermal tests, the smaller true $\Delta\varepsilon_{in}$ is observed in the 1050 °C half of the cycle, where the coating bears an insignificant load. Corrections for the 650 °C isothermal data improved agreement between the cyclic stress-strain behaviors of the bare and coated specimens. Still, corrections were relatively small and did not change any conclusions about the relative life behavior among the various test types.

Representative transmission electron micrographs of specimens failed in each test type are shown in figure 4. $\Delta\varepsilon_{in}$ was about 1×10^{-3} for the 650 °C isothermal test and both bithermal tests, but about 5×10^{-3} for the 1050 °C isothermal test. Dislocation densities are greater for the 650 °C test and both bithermal tests than for the 1050 °C test though it had a higher $\Delta\varepsilon_{in}$. However, in both bithermal tests most of the dislocations appear in interfacial networks around the γ' particles, as in the 1050 °C isothermal test. Only a few dislocations may be observed within the γ' particles for the bithermal tests.

Isothermal Fatigue Life

The isothermal fatigue life of bare and coated PWA 1480 at 650 and 1050 °C as a function of $\Delta\varepsilon_{in}$ is shown in figure 5. The true monotonic tensile ductilities for bare specimens tested at the two temperatures are plotted at one quarter cycle. Note first the bare specimen lives. Best fit lines have slopes of -0.58 and -0.50 for 650 and 1050 °C.

It may be seen in figure 5 that for these relatively short life tests there is no effect of the coating on the 1050 °C fatigue life. The longest test was only about 24 hr. However, at 650 °C the coating drastically reduces fatigue life for tests with small $\Delta\varepsilon_{in}$. For $\Delta\varepsilon_{in} \leq 10^{-4}$ coated specimens have a life one tenth to one hundreth that of bare specimens.

A sectioned coated specimen cycled at 650 °C to one quarter of its expected life for an $\Delta\varepsilon_{in}$ of 10^{-4}, 10^3 cycles, shows cracks in the coating already penetrating into the single crystal (fig. 6(a)). A bare single crystal exhibited a life of 10^5 cycles at a slightly higher $\Delta\varepsilon_{in}$. This bare specimen failed at an internal micropore, but the bare specimens tested at higher $\Delta\varepsilon_{in}$ failed at surface initiated cracks. In contrast, at 1050 °C both coated and bare single crystals failed more frequently by linkage of multiple cracks growing from internal pores, particularly at the lowest strain ranges (fig. 6(b)). Though coating cracks did develop, they initiated more slowly than at 650 °C for the same $\Delta\varepsilon_{in}$ and propagated more slowly in the single crystal.

In-Phase Bithermal Fatigue Life

Life for in-phase bithermal tests of bare and coated specimens on a $\Delta\varepsilon_{in}$ basis is shown relative to the isothermal results in figure 7. The in-phase

test life for bare specimens appears about the same as the 650 °C isothermal test life. However, this is somewhat misleading since bare specimens appear to fail as much due to gross oxidation as to fatigue for in-phase tests. A large volume of oxides spalled off the specimens in these tests. Significant oxide spalling did not occur in the 1050 °C isothermal tests, which does not strain the oxide at low temperature, or in the out-of-phase tests, which were much shorter lived, as will be seen. Further, compressive strains which cause the oxide scale to buckle and spall are small in the out-of-phase test (fig. 2).

Also, in the two load-controlled, lowest-strain in-phase tests of bare specimens, considerable necking occurred indicating significantly more ductility than for either 650 °C isothermal or out-of-phase tests. Thus, in the low-strain regime, actual failure due to fatigue in bare specimens in the in-phase test may be even more difficult than in isothermal 650 °C cycling. This view was confirmed by the long lives of coated specimens in low-strain in-phase tests (fig. 7). For $\Delta\varepsilon_{in}$ below about 2×10^{-4} the coating does not rapidly crack as in the 650 °C isothermal tests. A specimen cycled at a $\Delta\varepsilon_{in}$ of about 10^{-4} to the number of cycles that produced failure in the 650 °C isothermal test exhibited no coating cracking. Lives of coated specimens in the in-phase test were at least 10 times greater than in the 650 °C isothermal test at low strains.

The coated specimens which had not failed in about 10^4 cycles in the in-phase tests (fig. 7) were removed from testing and sectioned metallographically. It may be seen in figure 8(a) that some surface cracks had penetrated the single crystals. However, these cracks were very blunt relative to secondary cracks observed in specimens tested isothermally at 650 °C (fig. 6(a)). Internal cracks eminating from micropores, also apparent in figure 8(a), appeared more likely to lead to failure. A specimen tested to failure at a slightly higher strain range did fail from internal micropores (fig. 8(b)).

For the intermediate $\Delta\varepsilon_{in}$ tests shown in figure 7, life was drastically reduced for the in-phase tests of coated specimens falling near the life line for 650 °C tests of coated specimens. This suggests that at these strain ranges as in the 650 °C tests, early cracking in the coating may control life. Indeed, only a few small internal cracks were observed on fracture surfaces, though it was not clear whether the principle crack initiated in the coating or at the superalloy surface. Still, it appears that for the highest strain range tests, the in-phase and 650 °C test lives for both bare and coated specimens may all converge. This would be expected since in the limit all would fail in the first cycle at a $\Delta\varepsilon_{in}$ equal to the alloy ductility.

Out-Of-Phase Bithermal Fatigue Life

As shown in figure 9, both coated and bare specimens have short lives in the out-of-phase tests. Surface cracks initiate early as in the 650 °C tests of coated specimens. A test of a coated specimen at a $\Delta\varepsilon_{in}$ of 10^{-4} interrupted at one tenth of the expected life exhibited numerous cracks which had already penetrated into the single crystal. Figure 10 shows cracks initiated in the coating in out-of-phase tests on: (a) a longitudinal section, and (b) a fracture surface. It may be seen that these cracks appear much sharper than the cracks shown for in-phase tests in figure 8(a). However, as the results

for the bare single crystals show, the coating is not necessary for rapid crack initiation in the out-of-phase cycle. The lives of the coated and bare single crystals are indistinguishable in the strain range regime where they can be compared.

The detrimental effect of the out-of-phase cycle is actually much worse than is apparent in figure 9. Since many of the tests were conducted with $\Delta\epsilon_{in}$ below the resolution limit of about 10^{-4}, lives cannot be compared in that regime on a $\Delta\epsilon_{in}$ basis. Figure 11 shows life on the basis of maximum tensile stress, σ_{max}, in the cycle for the various bithermal tests and best fit curves for the isothermal test results. Here the full detrimental effect of the out-of-phase cycle in the long-life regime may be seen, and again it may be seen that bare as well as coated specimens suffer equally in the out-of-phase cycle.

Since crack initiation is observed to occur early in both the out-of-phase tests and 650 °C tests of coated specimens, the shorter life for the out-of-phase tests appears to be due to an increased crack growth rate. The detrimental effect of 1050 °C compressive half of the out-of-phase cycle appears to be environmental rather than mechanical. This is suggested by the results of out-of-phase tests of coated specimens conducted in vacuum shown in figure 12. It may be seen that in the absence of air the out-of-phase test lives are equal or greater than those for coated specimens tested at 650 °C.

DISCUSSION

Cyclic Stress-Strain Behavior

The deformation structures observed in specimens tested with similar $\Delta\epsilon_{in}$ in the isothermal and bithermal cycles (fig. 4) were in accordance with the observed cyclic stress responses (fig. 3). The 1050 °C stress responses of the in-phase and out-of-phase cycles are the same, both higher than that for the isothermal cycle. This correlates with the higher dislocation densities developed in the two bithermal cycles. These were, in fact, comparable to those for the 650 °C isothermal cycle, and correspondingly, the 650 °C stress response was similar among these three cycle types irrespective of the fact that for the bithermal cycles a greater fraction of the disloca- tions were in the γ-γ' interfaces. The somewhat lower stress for the in-phase cycle, which is in compression, may simply reflect the same tension-compression anisotropy observed in 650 °C isothermal tests. The 650 °C cyclic stress-strain behavior of this alloy is discussed in more detail elsewhere (ref. 11).

Isothermal Fatigue Life.

The isothermal fatigue behavior of the bare and coated superalloy single crystals (fig. 5) is straightforward. The lives of the bare crystals at both temperatures, as well as the coated crystals at 1050 °C, exhibit classic Manson-Coffin exponential dependencies on $\Delta\epsilon_{in}$ which extrapolate well to failure at the true monotonic tensile ductility at one quarter cycle. The slopes of -0.50 for the 1050 °C tests and -0.58 for the 650 °C tests of bare specimens are in the range of typical values for metals. The difference in

life between the two temperatures appears to be simply a result of the difference in ductility. Only the life of the bare crystal at 650 °C and the lowest $\Delta\varepsilon_{in}$ does not fit this description, appearing longer than expected. This specimen also failed in a different manner, from a large internal pore. At higher $\Delta\varepsilon_{in}$, cracks initiated at the surface and a defect was not always apparent. This suggests a lower limit in $\Delta\varepsilon_{in}$ necessary for surface crack initiation in the absence of a large defect. Such a transition from surface to internal failure with decreasing strain range is observed for other materials.

More rapid crack initiation in the coating than in the base crystals during 650 °C fatigue is due to the large $\Delta\varepsilon_{in}$ enforced in the coating by the stronger crystals. Fatigue behavior of the bulk PWA 276 NiCoCrAlY coating is actually superior on a $\Delta\varepsilon_{in}$ basis (ref. 8). The $\Delta\varepsilon_{in}$ required to produce failure in 10^4 cycles was 10 times greater for the coating than the PWA 1480 superalloy. However, the coating experiences the same $\Delta\varepsilon_t$ as the underlying crystal, and for small $\Delta\varepsilon_t$ which only strain the superalloy elastically, the coating can experience inelastic strain. For a $\Delta\varepsilon_t$ expected to produce failure in the superalloy in 10^4 cycles, the bulk coating had a life of only about 5×10^2. For higher $\Delta\varepsilon_{in}$, increased $\Delta\varepsilon_{in}$ in the superalloy greatly reduces its life, such that for a $\Delta\varepsilon_t$ of about 2×10^{-2}, both coating and superalloy have lives of about 10^2 cycles. Note in figure 5 that this is about the life at which the behavior of the coated and bare specimens begin to merge.

The fatigue study of the bulk NiCoCrAlY alloy showed that it has an enormous tolerance for inelastic strain at 1050 °C. Thus, even though the coating is very weak relative to the superalloy (fig. 3(a)), and thus suffers much larger $\Delta\varepsilon_{in}$, the life of the bulk coating was about 10 times greater than that of PWA 1480 over the range of $\Delta\varepsilon_t$ investigated, 8×10^{-3} to 2×10^{-2}. Thus, failures originating in the coating were not observed in tests at 1050 °C. And since for the strain ranges employed both the bare and coated specimens failed at internal pores, the same lives were observed. Further, this indicates little environmental damage to the bare specimens, at least in these tests which lasted less than a day.

In-Phase Bithermal Fatigue Life

In the high-strain regime, the in-phase bithermal fatigue lives appear to be controlled by the 650 °C ductility of the superalloy just as the 650 °C isothermal fatigue life (fig. 7). This may be somewhat surprising in that the 650 °C loading in the in-phase test is compressive, and the meaning of ductility in compression is not clear. However, failure in compression in the single crystal superalloy is not greatly different than in tension. Specimens loaded in compression at 650 °C will shear into two pieces along a single slip band at few percent strain just as in tension.

In the long-life regime, other mechanisms influence the in-phase test life. For bare specimens, oxidation has an obvious influence, producing considerable loss in load bearing area. Yet, the in-phase test life for bare specimens is not greatly foreshortened relative to the 650 °C isothermal fatigue life for bare specimens. The very blunt secondary cracks and evidence of considerable ductility, i.e., necking, relative to those for specimens

tested at 650 °C appear to indicate that crack growth during the 1050 °C tensile half of the in-phase cycle is more difficult than at 650 °C.

Behavior of coated specimens in the intermediate and long-life regime is more complex. In the intermediate life regime the in-phase test life of coated specimens may be controlled by coating cracking as in the 650 °C isothermal tests. This is suggested by the life similarity (fig. 7), and fractographic examination. However, it is clear the in the low-strain long-life regime crack initiation in the coating is more difficult in the in-phase test than in the 650 °C isothermal test. Life increases drastically for low-$\Delta\varepsilon_{in}$ in-phase tests, and the failure origin changes to cracks initiated at internal micropores. Further study of this apparently abrupt change in behavior for the in-phase test is required.

Out-Of-Phase Bithermal Fatigue

The surprise for out-of-phase tests is that lives of both bare and coated specimens are foreshortened by rapid initiation of surface cracks. Though for coated specimens, crack initiation may be due to the large enforced tensile strain in the coating at 650 °C just as for the 650 °C isothermal tests, rapid surface cracking also occurs in bare specimens. The similarity in life for coated and bare specimens suggests that crack initiation represents an insignificant fraction of life and that crack propagation in the base superalloy is the measure of life.

Since crack initiation is also rapid for 650 °C isothermal tests of coated specimens, the shorter life for the out-of-phase tests is probably due to more rapid crack propagation. That this is an environmental effect of the 1050 °C exposure is suggested by the increase in life for out-of-phase tests of coated specimens conducted in vacuum. These exhibited lives as long as the 650 °C isothermal tests of coated specimens. This finding and the rapid crack initiation in bare specimens for the out-of-phase test suggest a mechanism whereby a surface layer of oxide or dissolved oxygen forms during the 1050 °C half of the cycle and subsequently cracks brittlely during the 650 °C tensile loading. To be consistent one would have to postulate that this layer has more ductility at 1050 °C, and thus does not lead to rapid crack initiation or propagation in the cycles imposing tensile loading at 1050 °C, the isothermal or in-phase tests.

While the lives of bare and coated specimens are the same in out-of-phase tests in air it would be misleading to conclude that the coating has no effect on life in the out-of-phase test. It is reasonable to assume that if the coating did not fail, the superalloy beneath, protected from the environment, would have a much longer life. Even for the out-of-phase tests in vacuum rapid crack initiation probably occurs in the coating. In the 650 °C isothermal tests of bare specimens, where rapid surface crack initiation does not occur, lives are much longer than for the out-of-phase tests in vacuum. For the same reason, it is expected that out-of-phase test lives of bare specimens in vacuum would be much longer than those of coated specimens.

It may be somewhat surprising, that lives for the coated specimens in the out-of-phase tests in vacuum are not shorter than for those in the 650 °C isothermal tests. For the out-of-phase test, the tensile thermal mismatch strain

in the coating upon cooling makes $\Delta\varepsilon_t$ greater than the applied tensile strain. The coefficient of thermal expansion in the coating averages 1.8×10^{-6} C^{-6} greater than that of the superalloy over the temperature range of these tests (ref. 11). So, at most, the thermal strain in the coating upon cooling from 1050 to 650 °C could be about 7×10^{-4}, which would certainly be very damaging.

Two probable explanations come to mind. First, the coating is so weak at the higher temperatures in the cycle (ref. 6) that some of the developing thermal mismatch stress probably creeps out during cooling. Thus, less than the full thermal mismatch strain may remain as inelastic strain in the coating at the start of the 650 °C half of the cycle. Second, the first coating cracks develops so rapidly just under the influence of the applied strain ranges in the 650 °C isothermal tests conducted herein, that any acceleration due to the added thermal mismatch strain in the out-of-phase cycles is unnoticeable.

It should be recognized that the bithermal out-of-phase test is not only an idealized, but also a very severe test relative to what might be experienced in a gas turbine engine component because all the tensile strain is imposed in the temperature range where the materials have least ductility. The goal must be to eventually predict when component design and engine operating conditions will lead to premature failure in the coating rather than its intended function of environmental protection.

Further work is needed to confirm that rapid crack propagation and early crack initiation in bare specimens is largely an environmental effect in the out-of-phase test. There is, however, evidence that mechanical effects of the 1050 °C exposure, say creep damage, are not large. It has been shown that the 1050 °C creep-fatigue behavior of PWA 1480 can be explained by the $\Delta\varepsilon_{in}$ and the stress range without regard to the proportions of creep and rapid plastic strain (ref. 4). Also, the cyclic crack propagation rate of PWA 1480 has been shown to be insensitive to Cycle frequency by DeLuca and Cowles (ref. 9), albeit at 982 °C. The isothermal and bithermal fatigue behavior of the bare specimens observed herein can be rationalized based on the ductility of PWA 1480 at the two temperatures and a proposed environmental effect in the out-of-phase cycle.

It should be pointed out that cyclic crack propagation rates were found by DeLuca and Cowles to be the same for tests of PWA 1480 in more conventional out-of-phase and in-phase TMF cycles between 982 and 427 °C. These tests employed stress ratios approximating those developed in reversed strain fatigue tests such as those conducted herein. However, because of constraints on testing time, only crack growth rates greater than 10^{-5} mm per cycle were studied. It is reasonable that the environmental effect on the out-of-phase crack propagation suggested by the present study appears only in the near threshold regime where the crack advance per cycle is on the order of the depth of material which could be affected by oxidation during the 1050 °C half of the cycle. The bulk of crack propagation life in fatigue tests is spent in this regime, and thus any effect on propagation rate would be most noticeable.

Thermomechanical Fatigue Life

Finally, we should comment on the suitability of the simplified, bithermal test in understanding TMF behavior. The bithermal test shows the basic connection between TMF and isothermal fatigue behavior on a $\Delta\varepsilon_{in}$ basis in the high-strain regime where the basic mechanical damage processes in the superalloy control life. High-strain life is controlled by the ductility of the superalloy, and for the single crystal material this corresponds to the low temperature half of the bithermal cycle even when the loading is compressive. Further, this view of the basic TMF behavior of the superalloy provides a norm against which to judge the effects of the other damage mechanisms operating in the low-strain long-life regime, coating induced cracking and the environment. It should be pointed out however, that the simple correspondence between the low temperature isothermal and both bithermal cycles in the high-strain regime should not be expected in all materials. At the low temperature the single crystal superalloy has low ductility in compression as well as in tension, and the material is not greatly changed in the bithermal cycles relative to the microstructure developed during 650 °C isothermal cycling. This would not be the same for polycrystalline materials exhibiting grain boundary cavitation or alloys with less stable precipitates.

For coated specimens in the long life regime it is also necessary to consider the strain-temperature cycle in the coating since cracking in the coating can reduce the cycles necessary to develop a crack in the superalloy and exposure to the atmosphere may accelerate crack growth as shown for the out-of-phase cycle. Reasonable predictions of the cycles necessary to crack the coating in the 650 °C isothermal fatigue were obtained by estimating the $\Delta\varepsilon_{in}$ in the coating as shown in the appendix and using the fatigue life data for the bulk coating alloy (ref. 8). However, it was necessary to use only about 10 percent of the life to total separation, which corresponded to noticeable surface cracking in the bulk coating specimens. Such an approach to the more important task of predicting TMF life based on $\Delta\varepsilon_{in}$ in the coating will require more understanding of the constitutive behavior of the coating. Further, life prediction relevant to actual gas turbine engine components will be further complicated by the need to account for other damage mechanisms occurring in long time service, such as gross oxidation or hot corrosion, and interdiffusion between coating and superalloy.

The more complicated TMF cycles it is common the compare life based on $\Delta\varepsilon_t$ because of the difficulty in separating out $\Delta\varepsilon_{in}$. Success has been shown in using the applied $\Delta\varepsilon_t$ plus the thermal mismatch strain to correlate coating crack initiation for various coating-alloy combinations with different mismatches in the coefficient of thermal expansion (refs. 2 and 3). However, for the very different types of fatigue cycles studies herein, consideration of $\Delta\varepsilon_t$ alone does not contribute much to understanding. The lives for coated specimens in the two isothermal and two bithermal tests are shown on the basis of $\Delta\varepsilon_t$ in figure 13. It is difficult to see the connection between the two bithermal tests, the 650 °C isothermal test, and the 650 °C ductility demonstrated by considering the $\Delta\varepsilon_{in}$.

RESULTS AND CONCLUSIONS

Specimens of a single crystal superalloy, PWA 1480, both bare and coated with a NiCoCrAlY alloy, PWA 276, were tested in fatigue at 650 and 1050 °C and in TMF tests between the two temperatures. A "bithermal" TMF test type was employed in which mechanical loading was applied at the temperature endpoints. The test imposing tensile and compressive loading at the high and low temperatures, respectively, is termed "in-phase", and the opposite cycle is termed "out-of-phase". The following results and conclusions were obtained.

1. The only significant difference in cyclic stress response among the test types was a lower stress at 1050 °C for the isothermal test than for the bithermal tests. This correlated with an observed lower dislocation density than for the tests involving deformation at 650 °C.

2. Isothermal fatigue life on a $\Delta\varepsilon_{in}$ basis was as expected based on the monotonic tensile ductility of the superalloy. Tensile ductility and fatigue life are both much lower at 650 °C than at 1050 °C.

3. For high $\Delta\varepsilon_{in}$, lives for both in and out-of-phase bithermal tests approached that for the 650 °C isothermal tests. All are limited by the 650 °C ductility of the superalloy.

4. For low $\Delta\varepsilon_{in}$, the coating reduced life in 650 °C isothermal tests. Cracks initiated early in the coating and propagated into the superalloy.

5. Cracks initiated rapidly in bare as well as coated specimens in the out-of-phase tests and these propagated faster than in the 650 °C isothermal test. Tests in vacuum suggested this is an environmental effect due to the 1050 °C exposure and subsequent tensile straining at 650 °C.

6. For low $\Delta\varepsilon_{in}$, the coating improved life for in-phase tests.

7. The simplified bithermal TMF tests exhibit the same effects of more realistic TMF tests and show the connection with isothermal fatigue behavior in the high-$\Delta\varepsilon_{in}$ regime where the mechanical damage mechanisms in the superalloy control life. This view of basic TMF of the superalloy also provides a norm against which the effects of the other damage mechanisms operating in the low-$\Delta\varepsilon_{in}$, long-life regime, coating initiated cracking and the environment, may be assessed.

APPENDIX

In the 650 °C isothermal test and the 650 °C halves of the bithermal tests, the coating on the relatively small fatigue specimens used herein carries a non-negligable load. This appendix describes estimation of the actual stresses and strains in the coating and superalloy in the 650 °C isothermal tests. Since the coating carries very little load at 1050 °C, the actual strains in the superalloy are what is measured in the 1050 °C isothermal tests and the 1050 °C halves of the bithermal tests.

The elastic and inelastic strains in the superalloy and coating were calculated by requiring continuity and balance of forces. The elastic strains were calculated based on elastic moduli of 115 and 135 GPa, respectively for superalloy and coating at 650 °C. The inelastic strains were based on the cyclic stress-strain curves developed herein for the bare superalloy specimens and previously for the bulk coating (ref. 8). In an equation of the form

$$\sigma_{max} = a\Delta\varepsilon_{in}^{n'},$$

where σ_{max} is the maximum stress, a and n' were taken as 2900 and 0.15 for the superalloy, and 1350 and 0.193 for the coating. Equating the total strain in the superalloy and coating, and requiring the load on the superalloy to equal the applied load minus the load on the coating, the load on the coating is obtained for a give applied load. With these estimates of the true loads in the coating and superalloy, the corresponding estimates of the inelastic strains were obtained from the above cyclic stress-strain equations.

Since the coating experiences a greater inelastic strain than the superalloy, upon removal of, say, a tensile load the requirement of continuity drives the coating into compression and the superalloy is held in slight tension. By again requiring a balance of forces, the small elastic tensile strain in the superalloy at zero load may be calculated. This small elastic strain appears in addition to the true inelastic strain at zero load, making the apparent inelastic strain too large.

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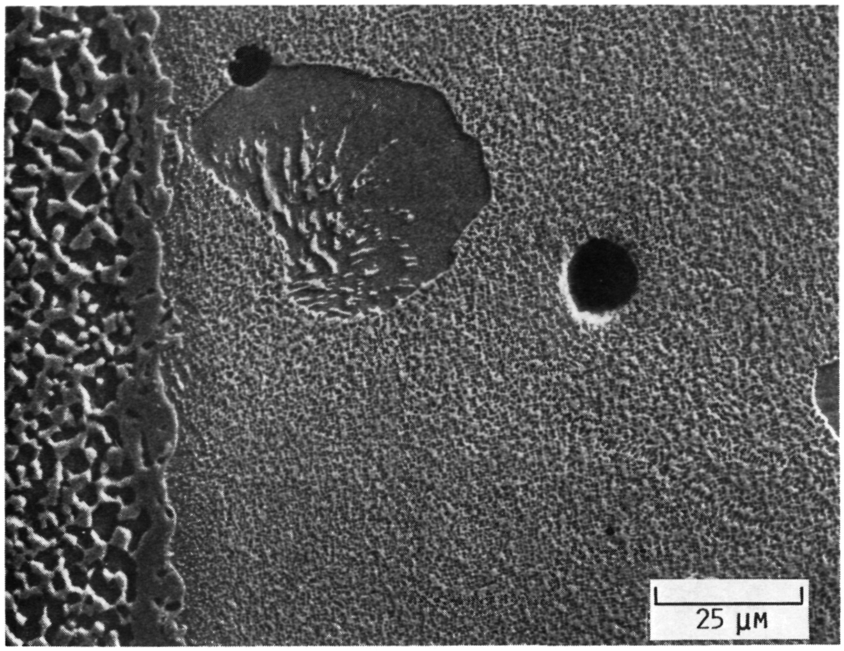


FIGURE 1. - MICROSTRUCTURES OF SINGLE CRYSTAL SUPERALLOY, PWA 1480, AND NiCoCrAlY OVERLAY COATING, PWA 276 (LEFT).

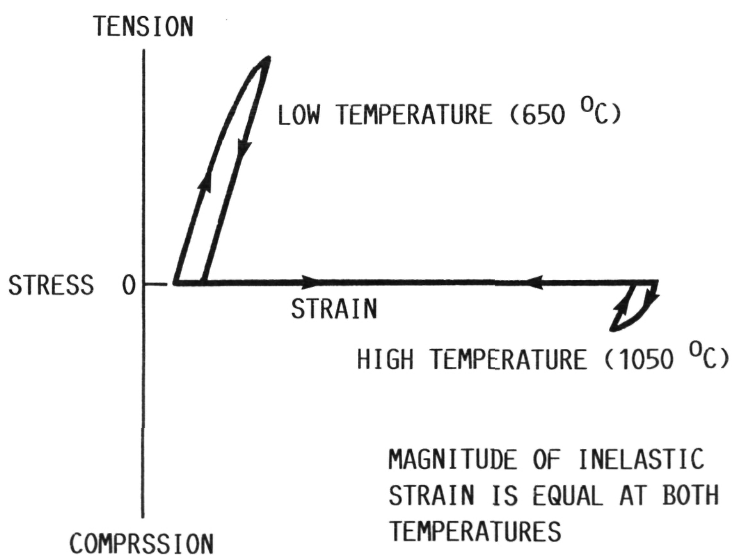


FIGURE 2. - STRESS-STRAIN HYSTERESIS LOOP FOR AN OUT-OF-PHASE BITHERMAL TEST.

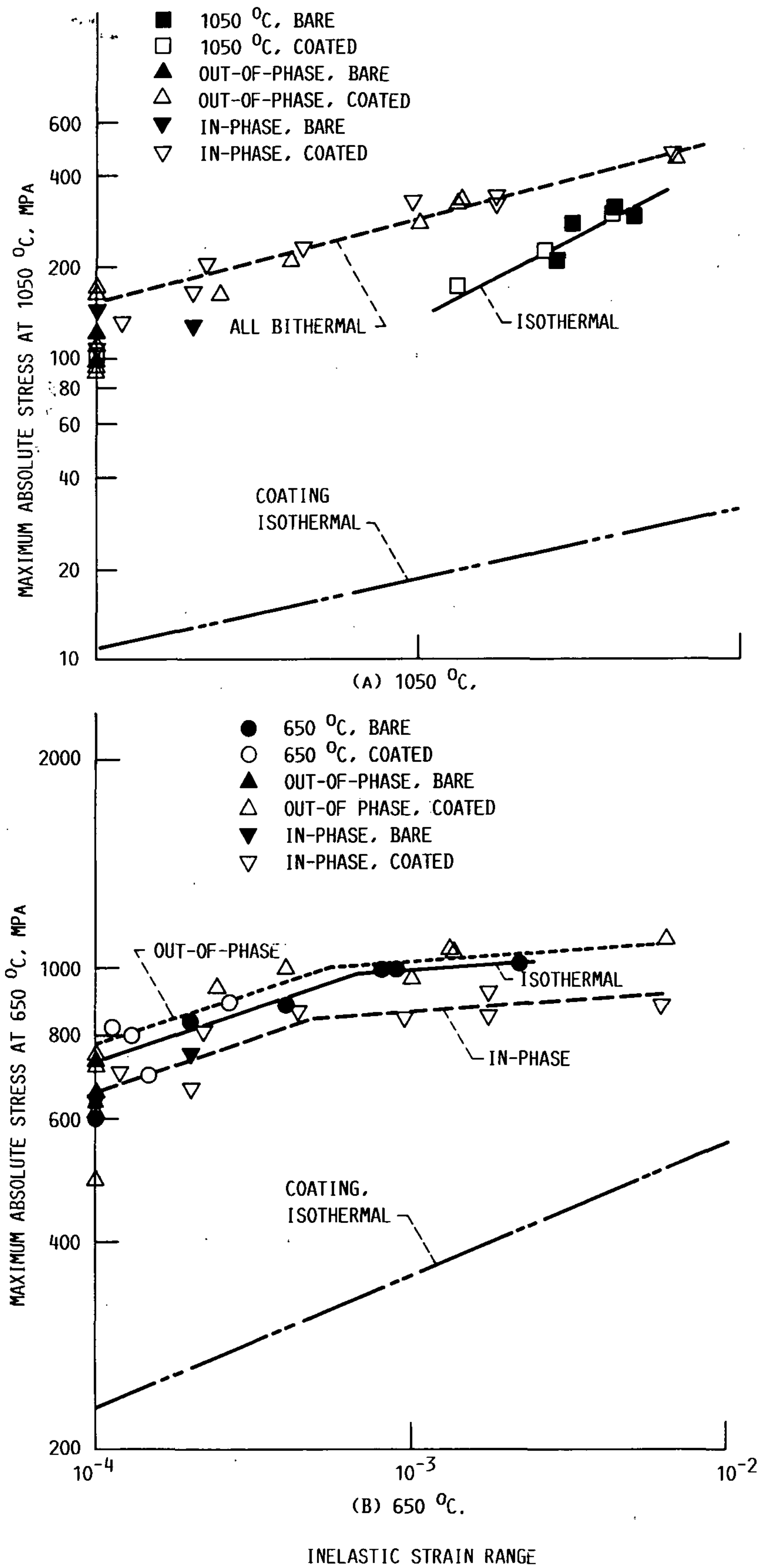
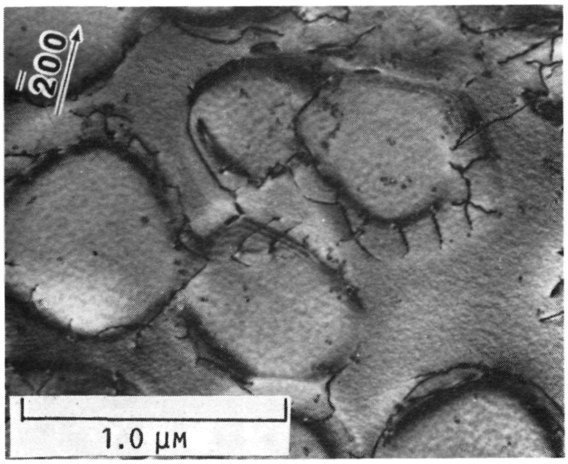
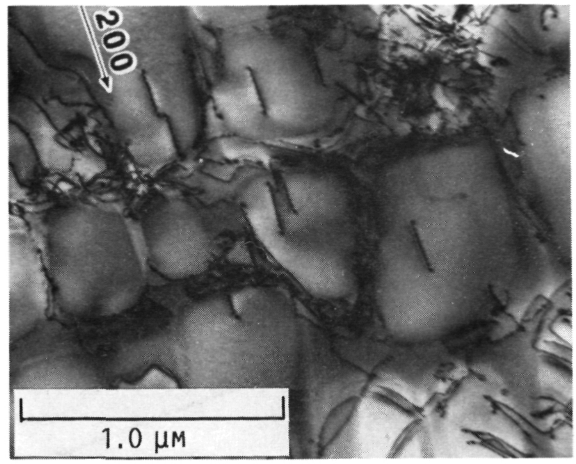


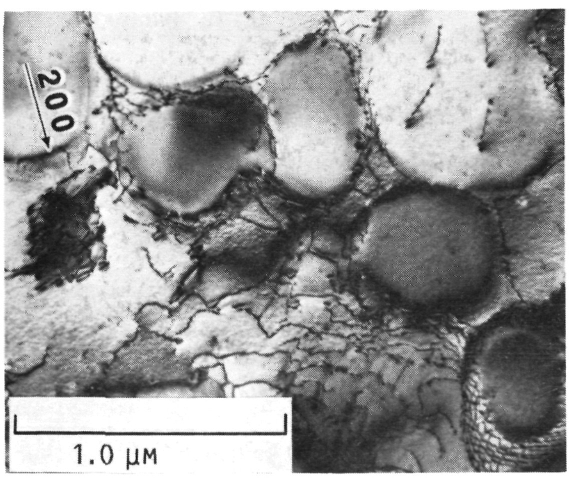
FIGURE 3. - MAXIMUM ABSOLUTE STRESS AT HALF LIFE AS A FUNCTION OF $\Delta\varepsilon_{IN}$.



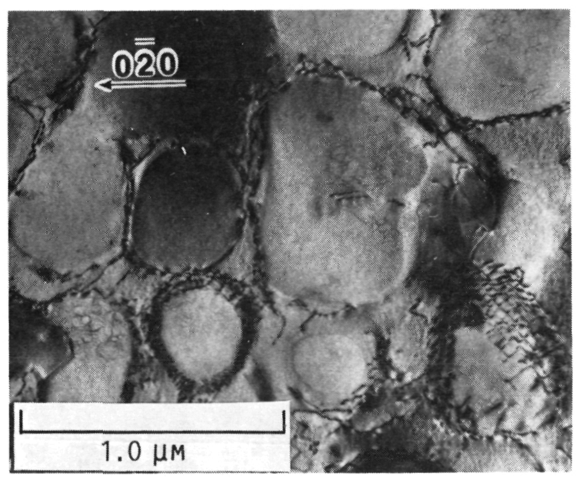
(A) 1050 ^{o}C.



(B) 650 ^{o}C.



(C) IN-PHASE BITHERMAL.



(D) OUT-OF-PHASE BITHERMAL.

FIGURE 4. - MICROSTRUCTURES OF SPECIMENS FAILED IN VARIOUS TESTS.

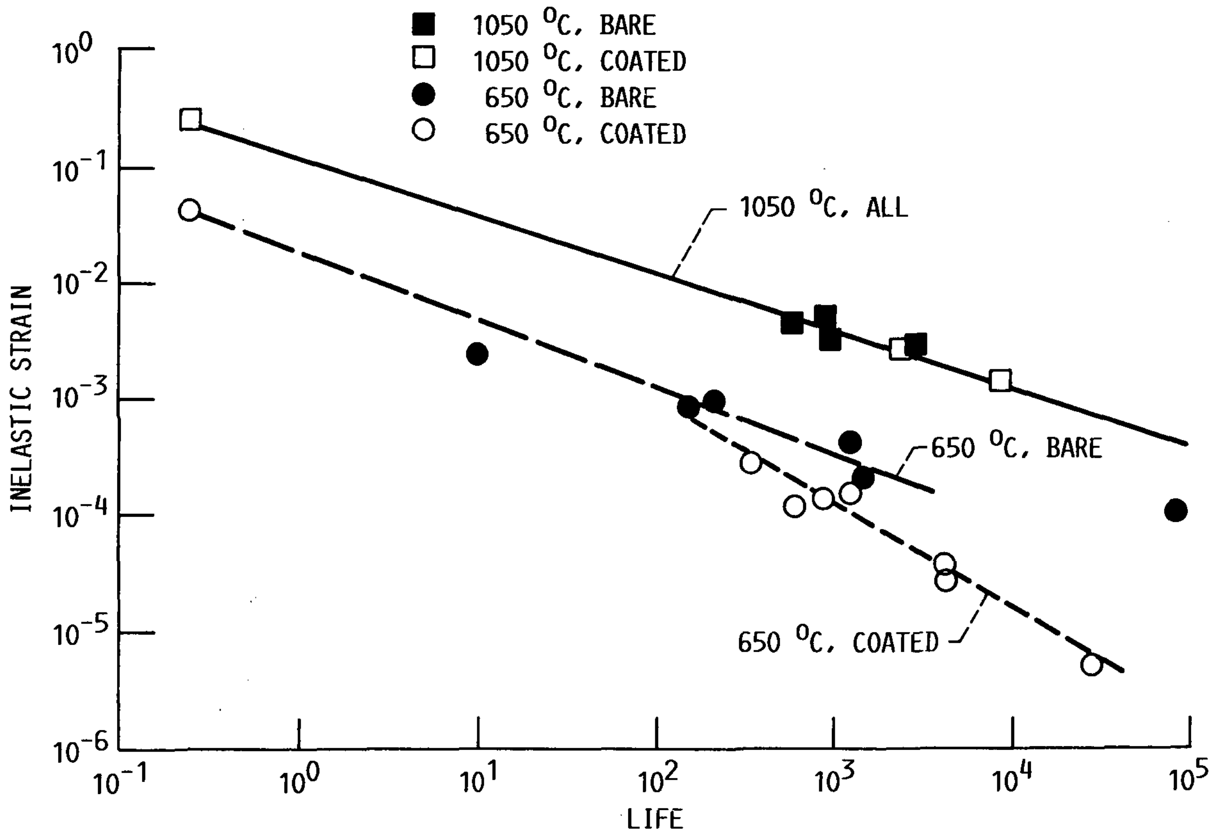
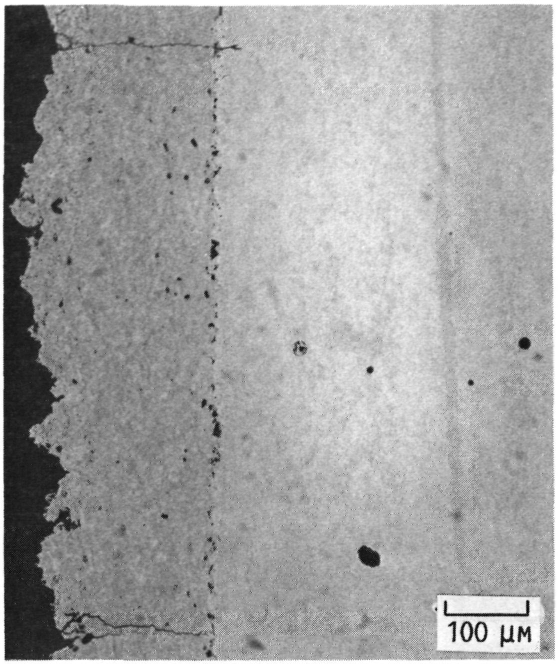
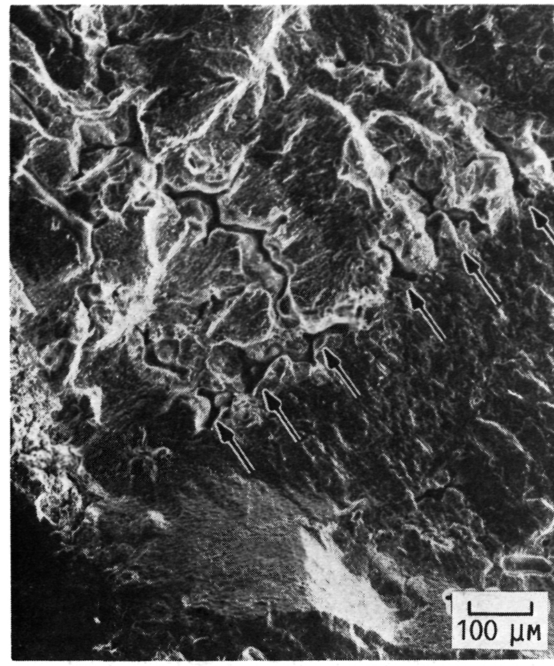


FIGURE 5. - FATIGUE LIFE-$\Delta\varepsilon_{IN}$ BEHAVIOR FOR BARE AND COATED SPECIMENS IN ISOTHERMAL TESTS.



(A) SURFACE CRACKS IN A 650 ^{o}C TESTS INTERRUPTED AT 250 CYCLES (1/4 OF LIFE).



(B) CRACKS INITIATED AT MICROPORES IN THE SUPERALLOY AT 1050 ^{o}C.

FIGURE 6. - FAILURE ORIGINS IN COATED SPECIMENS TESTED ISOTHERMALLY.

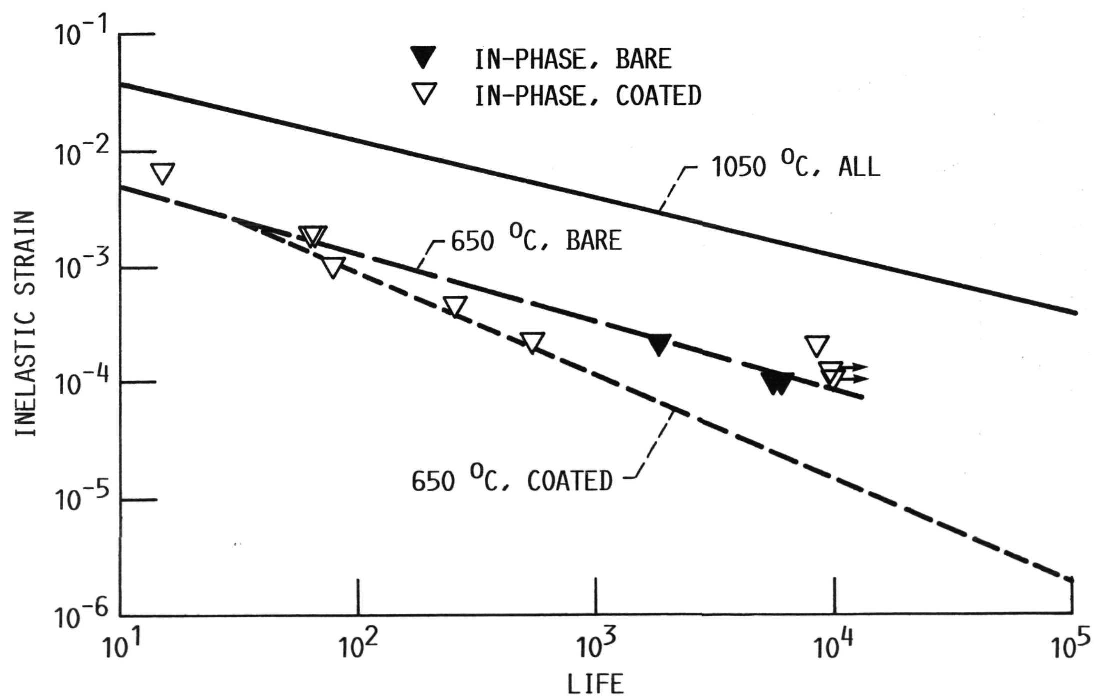
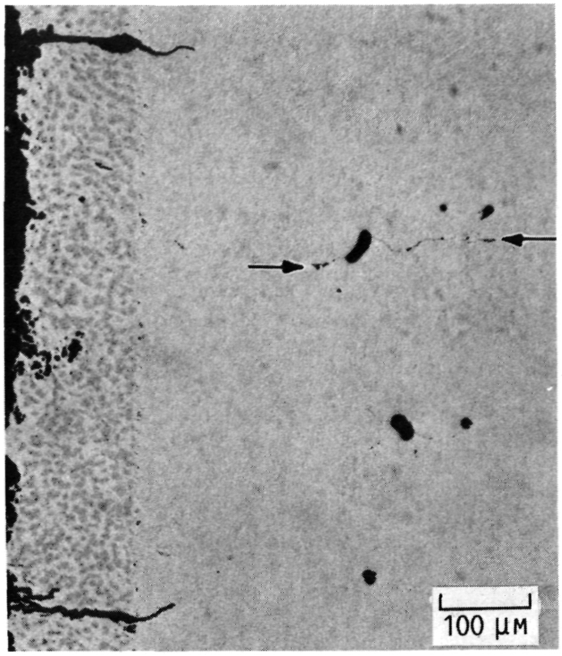
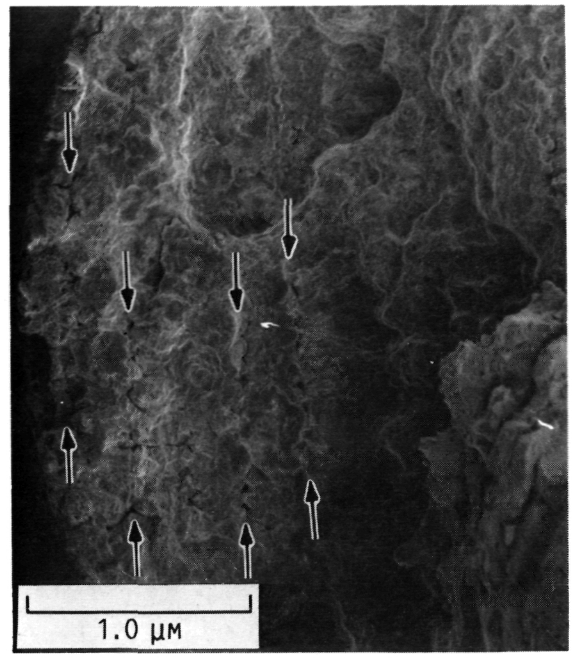


FIGURE 7. - FATIGUE LIFE-$\Delta\varepsilon_{IN.}$ BEHAVIOR FOR BARE AND COATED SPECIMENS IN IN-PHASE BITHERMAL TESTS RELATIVE TO ISOTHERMAL FATIGUE BEHAVIOR.



(A) BLUNT SURFACE CRACKS AND INTERNAL CRACK IN A TEST INTERRUPTED AT ABOUT 10^4 CYCLES.



(B) CRACKS INITIATED AT MICROPORES IN SPECIMEN FAILED AT SLIGHTLY LESS THAN 10^4 CYCLES.

FIGURE 8. - FAILURE ORIGINS IN COATED IN-PHASE TEST SPECIMENS.

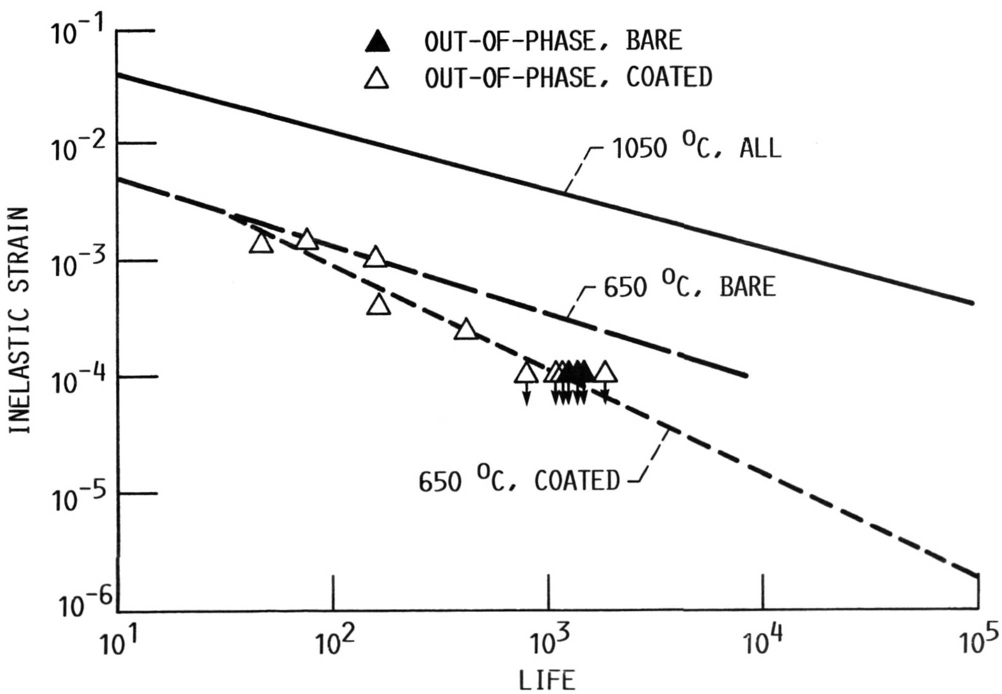


FIGURE 9. - FATIGUE LIFE-$\Delta\varepsilon_{IN}$ BEHAVIOR FOR BARE AND COATED SPECIMENS IN OUT-OF-PHASE BITHERMAL TESTS RELATIVE TO ISOTHERMAL FATIGUE BEHAVIOR.

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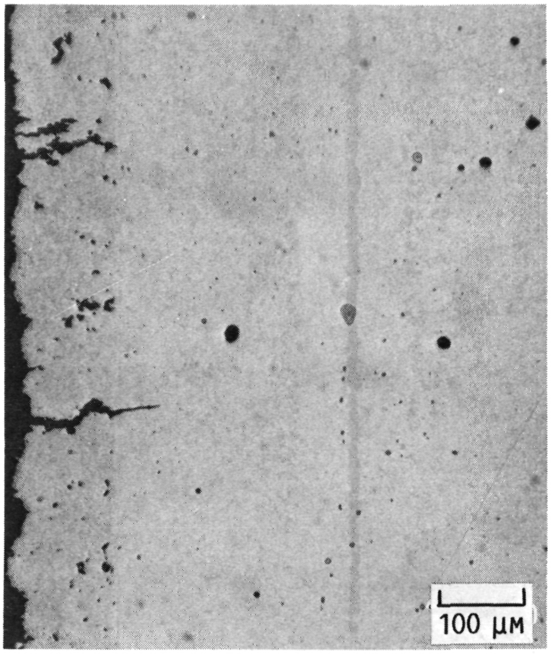
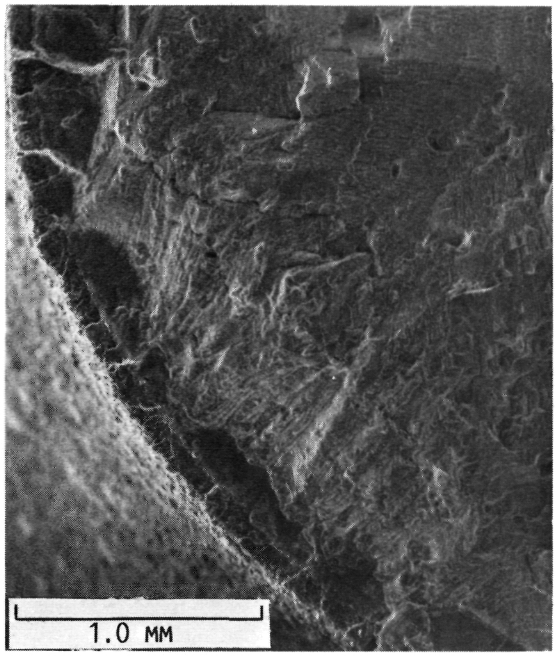


(A) SHARP CRACKS PENETRATING INTO SUPERALLOY IN A TEST INTERRUPTED AT 1/10TH OF EXPECTED LIFE.



(B) FRACTURE SURFACE SHOWING CRACKS INITIATED IN COATING.

FIGURE 10. - FAILURE ORIGINS IN COATED OUT-OF-PHASE BITHERMAL TEST SPECIMENS.

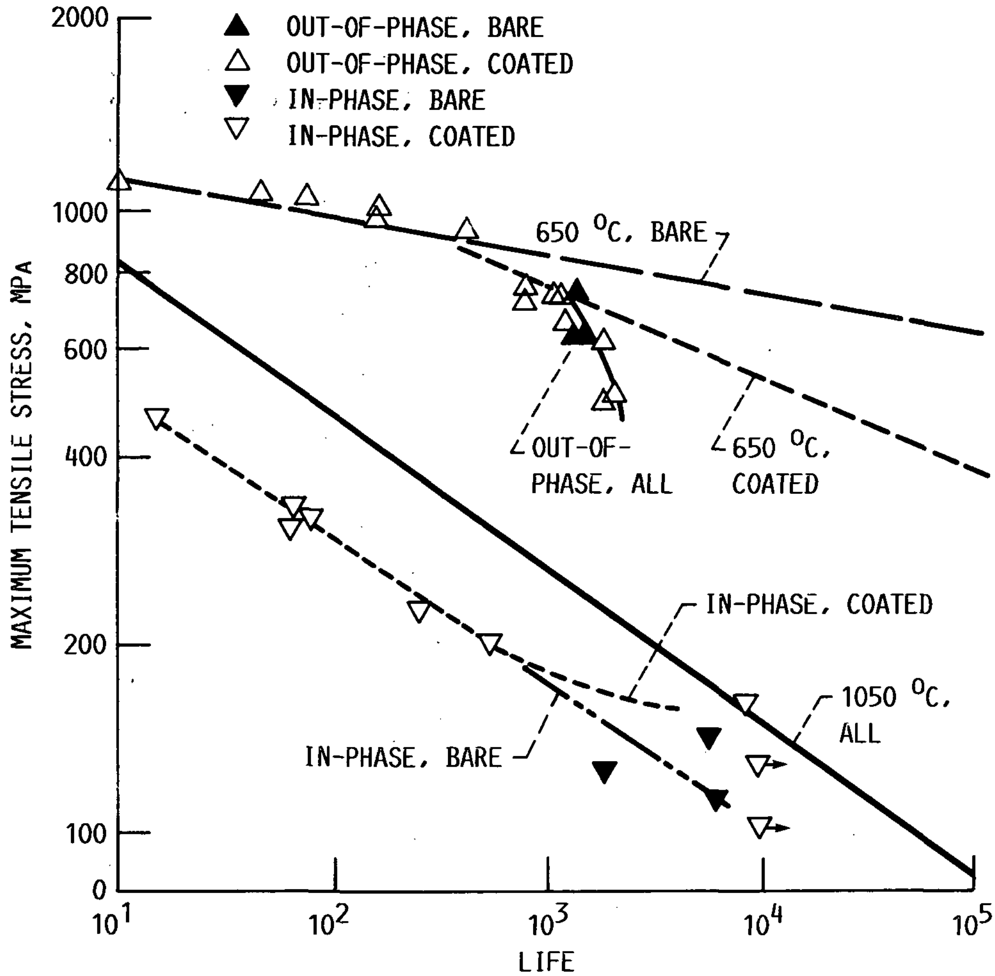


FIGURE 11. - FATIGUE LIFE-σ_{MAX} BEHAVIOR FOR BARE AND COATED SPECIMENS IN ALL TESTS.

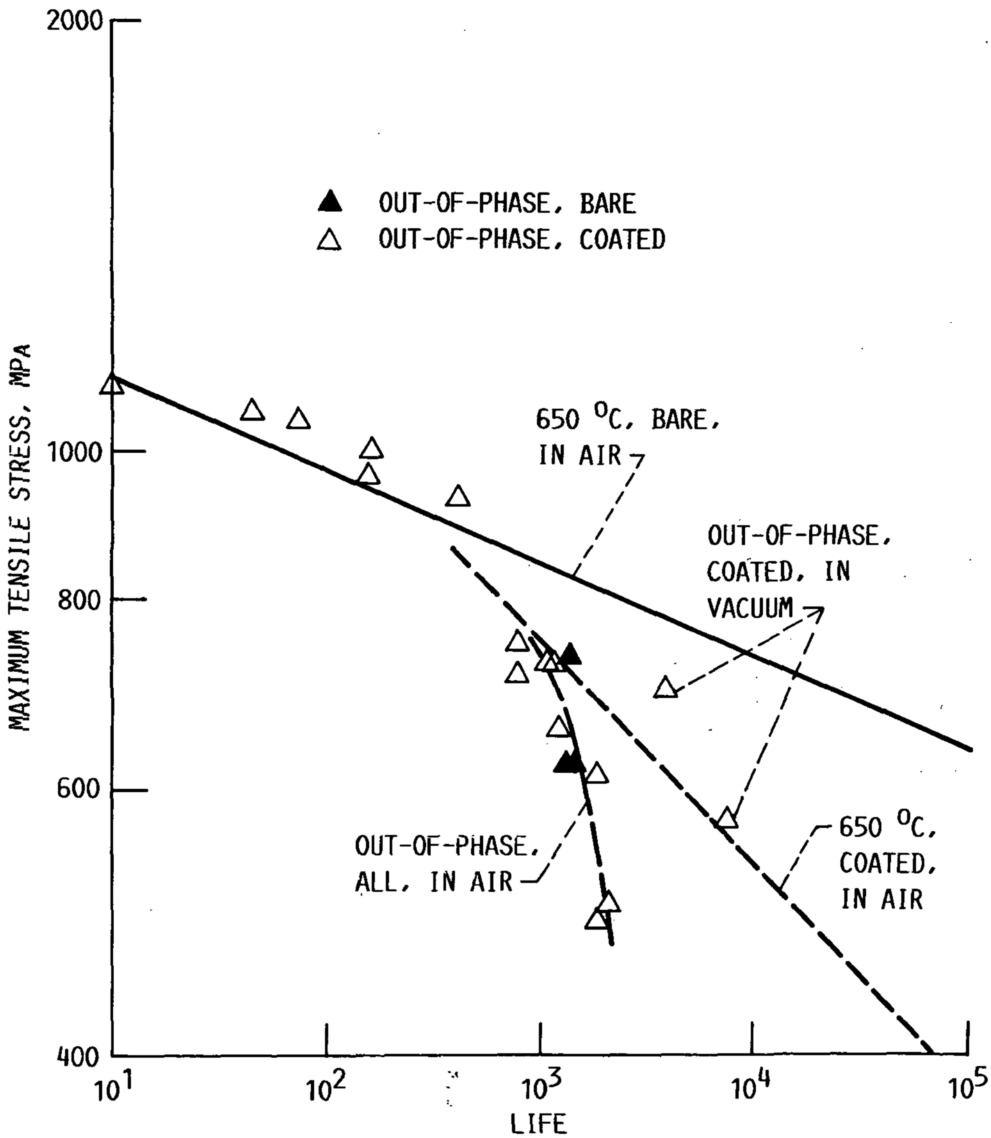


FIGURE 12. - RELATIVE FATIGUE LIFE-σ_{MAX} BEHAVIOR FOR COATED SPECIMENS IN OUT-OF-PHASE TESTS IN VACUUM.

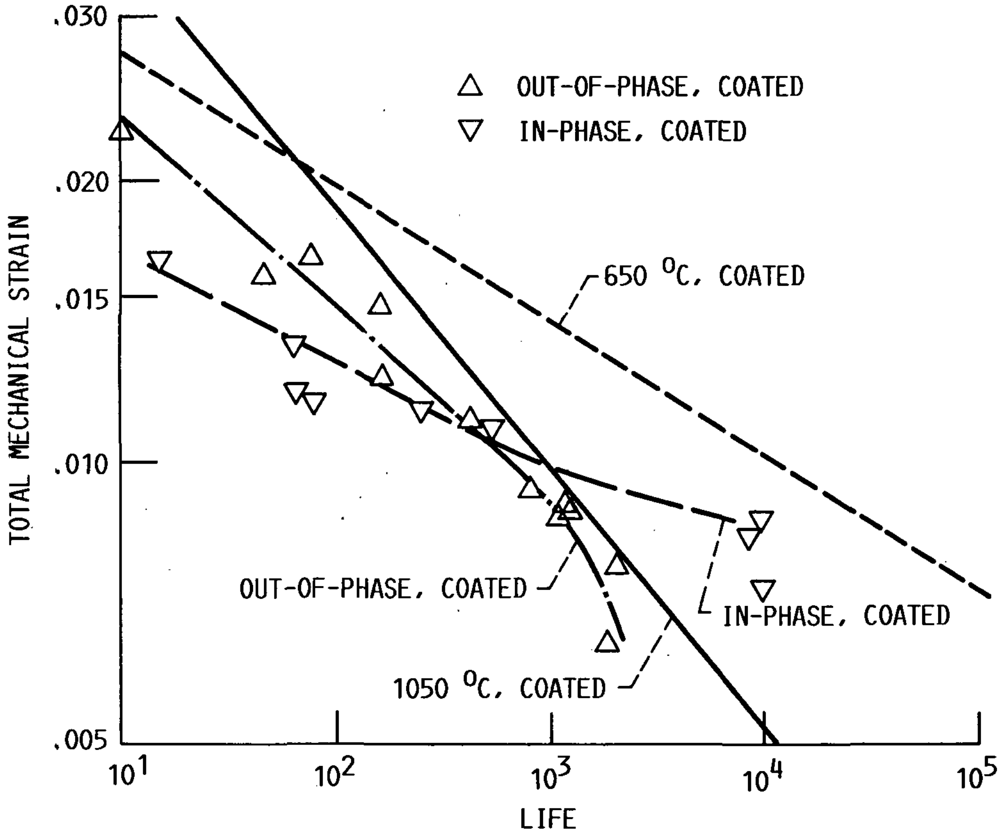


FIGURE 13. - FATIGUE LIFE-$\Delta\varepsilon_t$ BEHAVIOR FOR COATED SPECIMENS IN ALL TESTS.

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16. Abstract

Specimens of a single crystal superalloy, PWA 1480, both bare and coated with a NiCoCrAlY alloy, PWA 276, were tested in low-cycle fatigue at 650 and 1050 °C, and in "bithermal" thermomechanical fatigue tests. In the two bithermal test types, tensile strain was imposed at one of the two temperatures and reversed in compression at the other. In the high-strain regime, lives for both bithermal test types approached that for the 650 °C isothermal test on an inelastic strain basis, all being controlled by the low ductility of the superalloy at 650 °C. In the low-strain regime, coating cracking reduced life in the 650 °C isothermal test. The bithermal test imposing tension at 650 °C, termed "out-of-phase", also produced rapid surface cracking, but in both coated and bare specimens. Increased crack growth rates also occurred for the out-of-phase test. Increased lives in vacuum suggested that there is a large environmental contribution to damage in the out-of-phase test due to the 1050 °C exposure followed by tensile straining at the low temperature.

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